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RESEARCH MEMORANDUM

BEHAVIOR OF FORGED S-816 TURBINE BLADES IN STEADY-STATE OPERATION OF J33-9 TURBOJET ENGINE WITH STRESS-RUPTURE AND METALLOGRAPHIC EVALUATIONS

By F. B. Garrett, C. A. Gyorgak, and J. W. Weeton

Lewis Flight Propulsion Laboratory
Cleveland, Ohio

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RESEARCH MEMORANDUM

BEHAVIOR OF FORGED S-816 TURBINE BLADES IN STEADY-STATE OPERATION OF J33-9 TURBOJET ENGINE WITH STRESS-RUPTURE AND METALLOGRAPHIC EVALUATIONS

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SUMMARY

An investigation was conducted to determine the behavior of recently produced, forged S-816 turbine blades in a full-scale turbojet engine, and in particular, the scatter in performance of the alloy. The turbine blades were operated as continuously as possible at a temperature of 1500° F and a centrifugal stress of 21,500 pounds per square inch.

The operating lives of the turbine blades varied from 181 to 539 hours, a range of 358 hours. Stress-rupture properties of specimens cut from blade airfoils also varied considerably, as much as 1257 hours at 20,000 pounds per square inch and 1500° F. Since the variability or scatter of stress-rupture data is greater than that of blade performance, the scatter is probably caused by variations in the properties of the forged blades rather than by variations caused by engine operation or installation of the blades.

Metallographic examinations were made to determine possible causes of the scatter and although numerous differences in microstructures of blades were found, no consistent tendencies were observed and the findings did not permit an explanation of the scatter of blade performance. The results of the metallographic examinations and of the physical tests indirectly indicated that variables in fabricating methods caused the scatter in properties.

INTRODUCTION

The turbine blade is the component of the turbojet engine that must endure the most severe combination of stress, temperature, thermal shock, and vibratory conditions. Over a period of time there has been an increasing tendency on the part of blade manufacturers to use forged rather than cast blades.

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The present investigation was conducted to determine the operating lives of forged S-816 blades recently produced commercially and selected from U.S. Air Force stock, and to determine the variation, or scatter, in blade life in a full-scale engine. Differences in grain size, microstructure, or hardness were also investigated in order to determine their effect on variations in blade performance.

In this investigation, which was conducted at the NACA Lewis laboratory, a turbine rotor containing 46 forged S-816 blades was operated as continuously as was conveniently possible. Engine shutdowns were necessary, but intentional cycling (described in ref. 1) was avoided.

APPARATUS AND PROCEDURE

Turbine blades. - Commercially produced, forged S-816 (AMS 5765A) blades of the type required for the J33-9 turbojet engine were obtained from U.S. Air Force stock. The blades, fabricated during the 1949-50 period, were of the following nominal percentage composition:

Co	Ni	Cr	Mo	W	Cb+Ta	Fe	C
40 min.	19-21	19-21	3.5-4.5	3.5-4.5	3.5-4.5	5 max.	0.32-0.42

After forging, the blades were solution treated at 2150° F for 1 hour and water quenched; they were then aged at 1400° F for 16 hours and air-cooled.

The 46 blades selected at random for evaluation in the engine were radiographed for flaws, and the cross-sectional areas of three blades were measured with an optical comparator to enable calculations of distributions of centrifugal stress along the length of the blades by the method described in reference 2.

Engine operation. - Blades were installed in a J33-9 turbojet engine and operated to failure at rated speed under conditions as close to continuous operation as possible. The runs were conducted for a maximum of 16 hours per day after which time the engine was shut down. Further shutdowns were necessary to replace failed blades.

The J33-9 engine has 14 combustion chambers, employs a dual-entry centrifugal compressor, and develops a nominal thrust of 4000 pounds. More detailed descriptions of the engine, the fuel, and instrumentation are contained in references 1 and 3.

The engine was operated at 11,500 rpm, rated speed for the engine, to hold blade centrifugal stress and temperature constant at nominally 21,500 pounds per square inch and 1500° F, respectively, in the normal

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expected zone of failure, (2 in. above blade base). The blade stress was controlled by holding engine speed constant; the blade temperature was controlled by varying the engine exhaust-nozzle opening, which changed the gas temperature. Blade temperature was measured by thermocouples inserted in two special blades installed in the disk (ref. 4).

Terminology. - The term steady-state as used in this report is somewhat inaccurate because of the shutdowns previously described. However, compared with previous investigations at the Lewis laboratory, during which cycles of 20 minutes duration were used [15 min at rated speed and 5 min idle (ref. 1)] the term appears warranted. The term blade failure may be defined as either complete fracture or extensive cracking.

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Blade-elongation measurements. - Three blades were scribed near the trailing edge at 1/2-inch intervals (ref. 1). After each shutdown, the elongation of each scribed segment was measured with an optical extensometer having an accuracy of ±0.0001 inch. Accuracy of elongation measurements is, however, controlled by the degree of blade distortion and warpage; the elongation results should therefore be considered approximate rather than exact.

Macroexaminations of failed blades. - Visual examinations of all failures were made to determine, as nearly as possible, the manner by which the failures originated. Failures were then classified into the following categories (see ref. 1 for further details):

(1) Stress rupture: The origin of the fracture is characterized by an irregular granular surface having no evidence of fatigue-type failure.

(2) Fatigue: A smooth predominantly transgranular fracture surface occurs at the failure origin, sometimes showing the familiar concentric ring markings.

(3) Stress rupture followed by fatigue: The initiation of the fracture was similar to that described for the stress-rupture type failure and was followed by fatigue cracking. (Failures nucleated by intergranular oxide penetrations or by surface flaws that tear open under prolonged stress and which were followed by fatigue were considered within this category.)

In addition, measurements of the positions on the airfoil at which failures occurred were also made.

Metallographic examination of failed blades. - Metallographic examination was made of the blades that failed in the shortest and the second shortest times (181 and 192 hr, respectively), two intermediate blade

failures (255 and 317 hr), and two of the blades that failed after the longest operating times (434 and 539 hr). Fracture edges, areas believed to contain the origin of failures, and general microstructures were examined. Measurements were made of the smallest, largest, and most prevalent grain sizes of typical failed blades. The A.S.T.M grain size measurements are listed in groups of three digits throughout the report, the center figure representing the most prevalent grain size.

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Comparison of failed blades of present and previous investigations. - Prior to the present investigation, full-scale engine evaluations of nine alloys in a J33-9 engine were made; forged S-816 blades, older than those used in the present investigation, were included. Both cyclic (ref. 1) and steady-state (unpublished data) runs were made with the older-stock blades; the maximum steady-state operating life of the best S-816 blade was 78.9 hours. The best blade was examined, and micrographic comparisons were made with blades run in the present evaluation.

Stress-rupture strengths and other physical properties of blade airfoils. - Test specimens of rectangular cross section (1/8 by 3/16 in.) were machined from the airfoil portions of several blades. The shape of the specimen and the zone from which it was machined are shown in figure 1. Stress-rupture tests were run at 1500° F at various stresses; however, the majority of data were obtained at 20,000 pounds per square inch. Specimen temperatures were so controlled that there was less than a 5° F variation along the length of the specimen with the average temperature varying only ±5° F. Total elongation measurements were made after testing, and hardness measurements were made before testing.

Metallographic investigations of stress-rupture specimens. - Metallographic investigations were made of the fractured zones of several of the stress-rupture specimens. Included were examinations of fractured edges of test specimens and fragments of airfoils from which stress-rupture specimens were cut. In addition, grain-size measurements were also made of specimens metallographically examined.

RESULTS

Engine Operating Results and Blade Performance

Blade failures are plotted on the cumulative-frequency curve shown in figure 2. The first failure occurred in 181 hours, and the last failure in 539 hours, a range of 358 hours.

Classification of failures is as follows:

Failure	Percent
Stress rupture	9
Stress rupture followed by fatigue	80
Fatigue	2
Damage followed by fatigue	9

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The majority of the blade failures originated by the stress-rupture-followed-by-fatigue mechanism. By definition, these failures may be considered to have been initiated by a stress-rupture mechanism. Thus, a total of 89 percent of the failures originated by a stress-rupture mechanism.

Macroexaminations of Failed Blades

Photographs of an as-received blade and typical failed blades are shown in figure 3. The fractured blade which failed by stress rupture followed by fatigue is labeled to show the nucleus of the failure, the fatigue zone, and the last portion to fail. Stress-rupture cracks are present in the last portion to fail. The origins of failures are shown in figure 4. Of the 23 blades selected for macroexamination, 65 percent of the failures originated at the trailing edge, 35 percent at the leading edge, and 74 percent of the failure origins occurred in a zone between 2 to $2\frac{3}{8}$ inches from the blade roots. The average distance from the roots to the failure origins was 2.11 inches, and the centrifugal stress calculated for this distance was 21,500 pounds per square inch.

Elongation of Blades During Engine Operation

The solid curves of figure 5 show the elongation of three blades in the portion of the airfoil between $1\frac{3}{16}$ to $2\frac{3}{16}$ inches from the root; the dashed curves are similar data taken from reference 1. The blades of reference 1 were run in cyclic tests and failed in 44.7 and 49.4 hours. It is readily seen that the creep rates of the blades being investigated are lower than those of the previously investigated blades.

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Metallurgical Examination of Failed Blades

Grain size. - The specimens examined metallographically are listed in table I along with grain-size measurements. The blade that failed in the shortest time (181.2 hr) had a variable grain size near the fracture edge (fig. 6). The grain size in the immediate vicinity of the fracture was coarse near the center of the blade (A.S.T.M. 1-4-7) and fine at the trailing edge (A.S.T.M. 6-7-8). The blade that failed in the second shortest time (192 hr) had only slight differences in grain size while the remaining blades including those that failed in the longest periods were fine grained and relatively uniform in grain size.

Metallographic examinations of blade structure. - Photomicrographs showing areas at fracture edges and cracks in blades are presented in figure 7. In addition to the grain-size differences that are evident in figure 8, several other differences in the structures of the blades were observed. For example, the poorest blade (failed in 181 hr) contained more oxide penetrations, particularly at the tip of the trailing edge than did the other blades, (fig 8(c)). Such penetrations may have caused the shorter life relative to the other blades.

The edge conditions shown in figures 8(a) to 8(c) were evident in several of the blades, and the depletion zone shown in figure 8(d) was found in varying degrees in all blade segments examined. Examinations of as-received blades did not reveal this zone which is probably a decarburization zone formed during engine operation. The massive carbides were not visibly disturbed.

The failure zone of the blade that failed in 254 hours originated at the leading edge and, although figure 7 does not show it, there were several microvoids in this locality which may have nucleated the failure. None of the other blades contained such flaws. One of the best blades that was examined metallographically failed by cracking in 434 hours. Only a few intergranular cracks were visible at the trailing edge, although a small segregation of dirt (slag inclusions) and a slag stringer were found near the zone of the crack.

Metallographic examinations of fracture edges of the specimens that failed in 181, 254, and 317 hours confirmed that fatigue cracking took place over a portion of the fracture edges. The blades failed by stress rupture followed by fatigue. Fatigue was indicated by transcrystalline portions of the fracture edges.

Changes in blade microstructure with operating time. - The changes in the microstructure with increasing engine operating time are also shown in figure 7. An increase in carbide precipitation with time, particularly in slip lines and twin boundaries, and the subsequent agglomeration of these carbides appears to have occurred. This tendency is more readily visible in the stress-rupture specimens as will subsequently be described.

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Comparison of Blade Failures of Present and Previous Investigations

The microstructure of the blade previously run in steady-state operation (failure at 78.9 hr) was different in several respects from the structures of the blades being investigated (fig. 9). For example, except for one blade (the blade that failed in 181 hr) the grain size was more heterogeneous in the older alloy (investigated in ref. 1). In the older alloy the carbides were distributed in finer, more numerous dispersions; both salt and pepper carbides and massive carbides were more thickly distributed within the grains. The oxide penetrations in these blades seemed to be much deeper and more extensive than those found in the blades used in the present investigation. Figures 10(a) and 10(b) show, respectively, a very severe crack at the trailing edge and extensive oxide penetrations (possibly but probably not associated with cracks).

Stress-Rupture

The results of the stress-rupture tests at 1500° F are shown in table II. Stress-rupture lives at a stress of 20,000 pounds per square inch ranged from 89.5 to 1346 hours and precluded the possibility of plotting a reasonable stress-rupture curve. In addition, the elongation results and the hardness of the stress-rupture bars also showed no definite trends and could not be correlated with the stress-rupture strengths.

Metallurgical Examinations of Stress-Rupture Bars and Remnants of Blades from Which the Rupture Bars Were Cut

In an effort to determine the cause of scatter in stress-rupture lives, bars 1, 6, 9, and 11 and blade remnants from which they were cut were examined metallographically (see table II and fig. 11).

Grain size of fractured bars. - The grain size of the bar that did not fail after 1346 hours in the stress-rupture test was somewhat smaller than that of the other specimens (A.S.T.M. 6-7-8 against A.S.T.M. 4-6-8). Similar differences in grain sizes were found in the remnants of the blade airfoils from which the stress-rupture bars were cut (fig. 12).

Metallographic examination of stress-rupture bars. - The poorest and best stress-rupture bars (those that failed in 89.5 and 1346 hr) were carefully examined for dirt and porosity but no significant differences were found. The microstructures of all specimens which differed somewhat

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because of various times at a test temperature of 1500° F also differed in the size of primary carbides and the number of primary carbides per unit of volume; for example, more carbides were present in the specimen that failed in 1346 hours than were present in the specimen that failed in 89.5 hours. Whether this difference is significant could not be determined except on a statistical basis (which is extremely time consuming) and the value of such a determination is questionable.

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The photomicrographs of figure 11 were taken at or near the failure zones of the rupture bars. Cracks, intergranular tears, and fracture edges are evident.

Changes in microstructure of stress-rupture specimens with time at temperature of 1500° F. - If the structures of the untested specimens shown in figure 12 and the structures shown in figure 11 are compared, it may be seen that, as the time at the test temperature increases, the precipitation within the grains, in slip lines and in twin boundaries, increases until at the longer times these precipitates tend to spheroidize. Furthermore, the matrices do not become spotted with salt and pepper precipitates such as those found in the older alloy (fig. 9). It is, of course, possible that the large number of agglomerated carbides etches rapidly and prevents subsequent satisfactory etching of the matrix carbides, although this is not believed to be the case.

Additional metallographic information not fully revealed by photomicrographs shown. - The fracture edges of the stress-rupture bars were intergranular and the usual intergranular tears below the fracture zone were also present. Twin and deformation bands, containing no visible precipitation, were observed under oblique light on a specimen that was cut from the same airfoil as one of the stress-rupture bars. Such deformation effects could have been caused by the water quenching that followed the solution treatment given the blades after forging. The aging treatment of 16 hours at 1400° F produced no visible precipitation within the twin bands and slip lines; in fact, figure 11(a) shows that after an additional aging treatment of 89.5 hours at 1500° F, relatively little precipitation occurred in these areas. When the specimens were tested for longer times at 1500° F, extensive precipitation occurred in these planes as evidenced by several photomicrographs shown in figure 11.

DISCUSSION OF RESULTS

The first blade failure did not occur until after 181 hours of steady-state engine operation at a blade temperature of 1500° F and a blade stress of 21,500 pounds per square inch. A first failure time of such magnitude compares favorably with stress-rupture data available for S-816 bar stock (the 100 hr strength is known to be 23,000 lb/sq in. at 1500° F) and, since almost all the blade failures were initiated by stress-rupture, it

is evident that properties of the forged blades were equal to the properties of the bar stock; a relation that has not often been observed for blade alloys in the past. Furthermore, the first failure time of 181 hours was a considerable improvement over the first failure time of 41 hours obtained under similar engine operating conditions with S-816 blades forged several years ago (table III).

Results of cyclic operation of the older and of the more recently forged blades are also shown in table III, and again the more recently obtained blades are superior to the older blades as is shown by the first failure times of 163 and 33.6 hours, respectively.

Perhaps as important as the first failure times of the blades, are the scatter of blade lives and of stress-rupture results. It was shown that blade lives varied from 181 to 539 hours, a range of 358 hours, and that stress-rupture lives at 1500° F and 20,000 pounds per square inch varied from 89.5 to 1346.3 hours, a range of 1257 hours. A range of such magnitude prevents correlation of blade performance to stress-rupture properties, elongation, and hardness results. The cumulative-frequency curve (fig. 2) shows that once the first failure has occurred, other failures will soon follow. Thus, the replacement of a failed blade in service by a new blade is relatively useless, unless the first failure has occurred in an abnormally short time and is not typical of the normal behavior of the alloy.

Microstructural differences in grain sizes, degree of oxide penetrations, carbide particle sizes, and inclusions were observed in specimens cut from turbine blades and stress-rupture bars. Some of these differences may have been significant, but since no consistent metallographic tendencies could be found, it is impossible to explain the scatter of physical properties. More exhaustive and fundamental investigations could be made of the blades evaluated, as well as of the poorer blades previously investigated and used as a basis of comparison. However, such efforts were not considered warranted after the metallographic investigations presented in this report indicated no trends of significance and indirectly showed that a complete history of the blade manufacturing procedures would be necessary to determine concretely the causes for scatter of blade life and stress-rupture properties.

Because of the differences in stress-rupture strength, and because only one turbine blade failed in fatigue (the last to fail) the scatter of blade performance cannot possibly be attributed to engine variables, or changing blade failure mechanisms. The unusually large scatter of stress-rupture properties graphically shows that variables in fabricating methods existed.

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Microstructural changes were observed in specimens run in the engine and specimens tested in stress-rupture at 1500° F for varying periods of time. Examination of untested segments of blade airfoils showed that the heat treatment given the alloy after forging, including an aging of 16 hours at 1400° F, did not cause precipitation of carbides in twin or slip boundaries. The stress-rupture specimen that failed after 89.5 additional hours at 1500° F contained very little precipitation. With increasing time at 1500° F more visible precipitation occurred and a tendency to spheroidize was obesrved. Agglomeration of many of the carbides occurred in the relatively short time of 181 hours (fig. 5(a)) but weakening of the alloy apparently did not occur as a result of the agglomeration; this was shown by the fact that both blades and stress-rupture bars withstood several hundred additional hours of stress at 1500° F after the initial agglomeration of some carbide particles.

A similar phenomenon was found to occur in an unreported investigation at the Lewis laboratory that involved the heat treatment and stress-rupture testing of wrought Haynes-Stellite 21 (AMS 5385). In this case, structures of specimens that were solution treated at 2250° F and aged at 1950° F for 72 hours were agglomerated and spheroidized to an appreciable degree prior to stress-rupture testing. The specimens proved to be of greater high-temperature strength than many specimens heat-treated at lower aging temperatures that did not contain agglomerated carbides.

SUMMARY OF RESULTS

The following results were obtained from the operation of forged S-816 turbine blades run in a J33-9 turbojet engine as continuously as possible. Such engine runs, termed steady-state runs, were conducted using a rotor speed of 11,500 rpm that produced a centrifugal stress of 21,500 pounds per square inch in the midsection of the blades, at a blade temperature of 1500° F. In addition, results of stress-rupture tests and metallographic investigations are presented.

1. The operating lives of the forged S-816 turbine blades evaluated varied from 181 to 539 hours. The time required for the first blade failure was of the same order of magnitude as the published stress-rupture strength of bars of S-816.

2. Stress-rupture tests at a temperature of 1500° F and a stress of 20,000 pounds per square inch of specimens cut from blade airfoils showed a range of rupture lives of from 89.5 to 1346 hours.

3. Ranges in blade lives of 358 hours and in stress-rupture lives of 1257 hours precluded the correlation of stress-rupture lives to blade performance. The large scatter of stress-rupture properties indicated that variables in fabrication methods existed.

4. Of the blades that failed in the engine, 23 were selected for macro-examination; 65 percent of the failures originated at the trailing edges, 35 percent originated at the leading edges, and 74 percent of the failure origins occurred in a zone between 2 to $2\frac{3}{8}$ inches from the blade roots. The average distance from the roots to the failure edges was found to be 2.11 inches.

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5. Differences in grain sizes, degree of oxide penetrations, size of carbide particles, and presence of inclusions, as well as other metallographic differences, were found in specimens cut from the turbine blades and from fracture areas of stress-rupture bars. However, no consistent differences between specimens other than increasing precipitation and agglomeration of carbides with time at 1500° F could be found, and correlations of structures with blade or stress-rupture lives were not feasible.

6. Agglomeration or spheroidization of some of the carbides formed by precipitation during aging or during engine operation at 1500° F was observed in the blade that failed in 181 hours, and in a stress-rupture specimen that failed in 219 hours. Both blades and stress-rupture specimens withstood several hundred additional hours of stress at a temperature of 1500° F after the initial agglomeration. Thus, agglomeration or spheroidization of precipitates, normally thought to cause a decrease in mechanical properties, does not appear, in itself, evidence of high-temperature weakness.

7. A previously run S-816 turbine blade was examined and compared with the blades of the present investigation. This blade was the best of several blades run; it failed in 78.9 hours, which is appreciably lower than the minimum failure time obtained in the present investigation. Microstructural differences between the older and newer blades were found, and these differences may explain the superior behavior of the newer alloy. For example, the structures of the newer blades contained fewer visible precipitates than the structure of the older blade. Furthermore, the grain size of the newer alloy was finer and more uniform, and oxide penetrations were less extensive.

CONCLUDING REMARKS

The stress-rupture data of forged S-816 turbine blades investigated herein would indicate that the blades should have operating lives of more than 100 hours at a stress of 20,000 to 22,000 pounds per square inch at a temperature of 1500° F. Thus, the performance of the blades in this investigation compared more favorably with the expected behavior of the alloy than blades in previous performance tests. However, the scatter in blade performance and, in particular, stress-rupture lives and the

subsequent unsuccessful attempts to correlate blade performance with hardness, elongation, or metallographic results leads to the conclusion that the effects of such factors as fabricating variables, which cannot always be observed in the laboratory, account for the variability in life.

It should also be observed that the cumulative-frequency distribution of the blade failures would indicate that replacement of failed blades in service, unless the failure has occurred in abnormally short times, would be relatively useless for additional failures would be expected in very short times.

Lewis Flight Propulsion Laboratory
National Advisory Committee for Aeronautics
Cleveland, Ohio

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TABLE I. - GRAIN-SIZE MEASUREMENTS AND HARDNESS OF FAILED BLADES

Blade	Failure time, hr	A.S.T.M. grain size[a] and location of measurement		Type of failure	Rockwell C hardness[b] (converted from Rockwell A)
1	181.2	1-4-7	Near center of fracture edge	Stress rupture followed by fatigue	28
		6-7-8	Near fracture at trailing edge		
		5-7-8	General area at trailing edge		
2	192	6-7-8	Leading-edge cross section near fracture	Stress rupture	30.4
		5-6-7	Center cross section near fracture		
		6-7-8	Trailing-edge cross section near fracture		
3	254.6	6-7-8	Near fracture at leading edge	Stress rupture followed by fatigue	32
		6-7-8	General area at leading edge		
4	316.7	5-7-8	Near fracture at trailing edge	Stress rupture followed by fatigue	32.4
		6-7-8	General area at trailing edge		
5	434.2	6-7-8	Leading-edge cross section	Stress rupture (cracked not fractured)	31.4
		6-7-8	Trailing-edge cross section		
6	538.8	6-7-8	Leading-edge cross section near fracture	Stress rupture	31.8
		5-7-8	Center cross section near fracture		
		6-7-8	Trailing-edge cross section near fracture		

[a]Extreme values of grain size are presented along with most prevalent grain size which is listed in center of column.

[b]Average of at least five readings.

TABLE II. - STRESS-RUPTURE RESULTS, HARDNESS, AND GRAIN-SIZE MEASUREMENTS OF SPECIMENS CUT FROM BLADE AIRFOILS

Specimen	Stress, lb/sq in.	Rupture life, hr	Elongation in $1\frac{1}{2}$-inch gage length, percent	Rockwell C-Hardness[a] (converted from Rockwell A)		A.S.T.M. grain size[b]
				Stress-rupture specimen	Remanents[c] of airfoils	
				Before test	Not tested	
14	38,000	2.5	32	25.5	25.8	
13	35,000	6.8	25.3	25.0	25.6	
12	30,000	9.5	18.0	26.0	24.4	
11	20,000	89.5	12.5	25.2	26.6	4-6-8
8	20,000	135.6	23	----	26.4	
4	20,000	162.0	18.7	----	25.6	
7	20,000	179.4	25	----	21.6	
10	20,000	184.2	15.6	25.5	30.0	
9	20,000	218.8	18	25.0	27.0	4-6-8
5	20,000	545.4	31.2	----	26.2	
6	20,000	618.0	22.9	25.0	26.0	4-5-8
1	20,000	1346.3[d]	1.5 - 2[d]	----	32.4	6-7-8

[a]Average of at least five readings.

[b]Extreme values of grain size are presented along with most prevalent grain size listed in center of column.

[c]From leading and trailing edge remnants of blades.

[d]Test bar didn't fail; test discontinued.

TABLE III. - BLADE PERFORMANCE OF S-816 ALLOY FROM OTHER BLADE EVALUATION PROGRAMS INCLUDED FOR COMPARATIVE PURPOSES

Order of blade failure	Cyclic evaluations[a] of blade life of older alloy, hr	Cyclic evaluations[b] of blade life of older alloy, hr	Cyclic evaluations[b] of blade life of newer alloy, hr
1	33.6	41	163
2	34.8	50.5	253
3	44.1	54.1	286
4	44.7	55.1	291
5	44.7	58.7	291
6	49.4	78.9	292
7	----	----	294
8	----	----	296
9	----	----	302
10	----	----	306
11	----	----	307
12	----	----	394

[a]Reference 1.

[b]Unpublished data.

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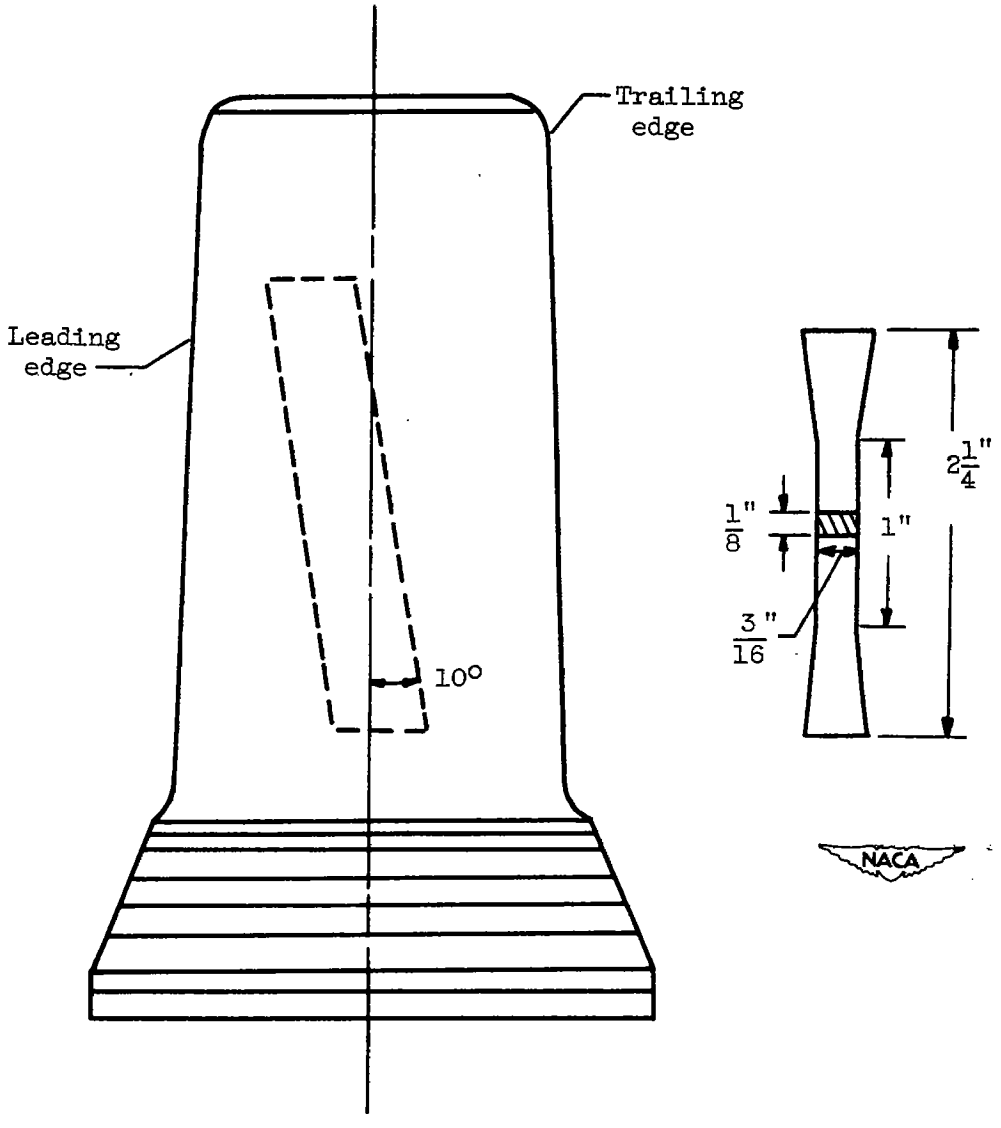


Figure 1. - Blade stress-rupture specimen and zone from which it was machined.

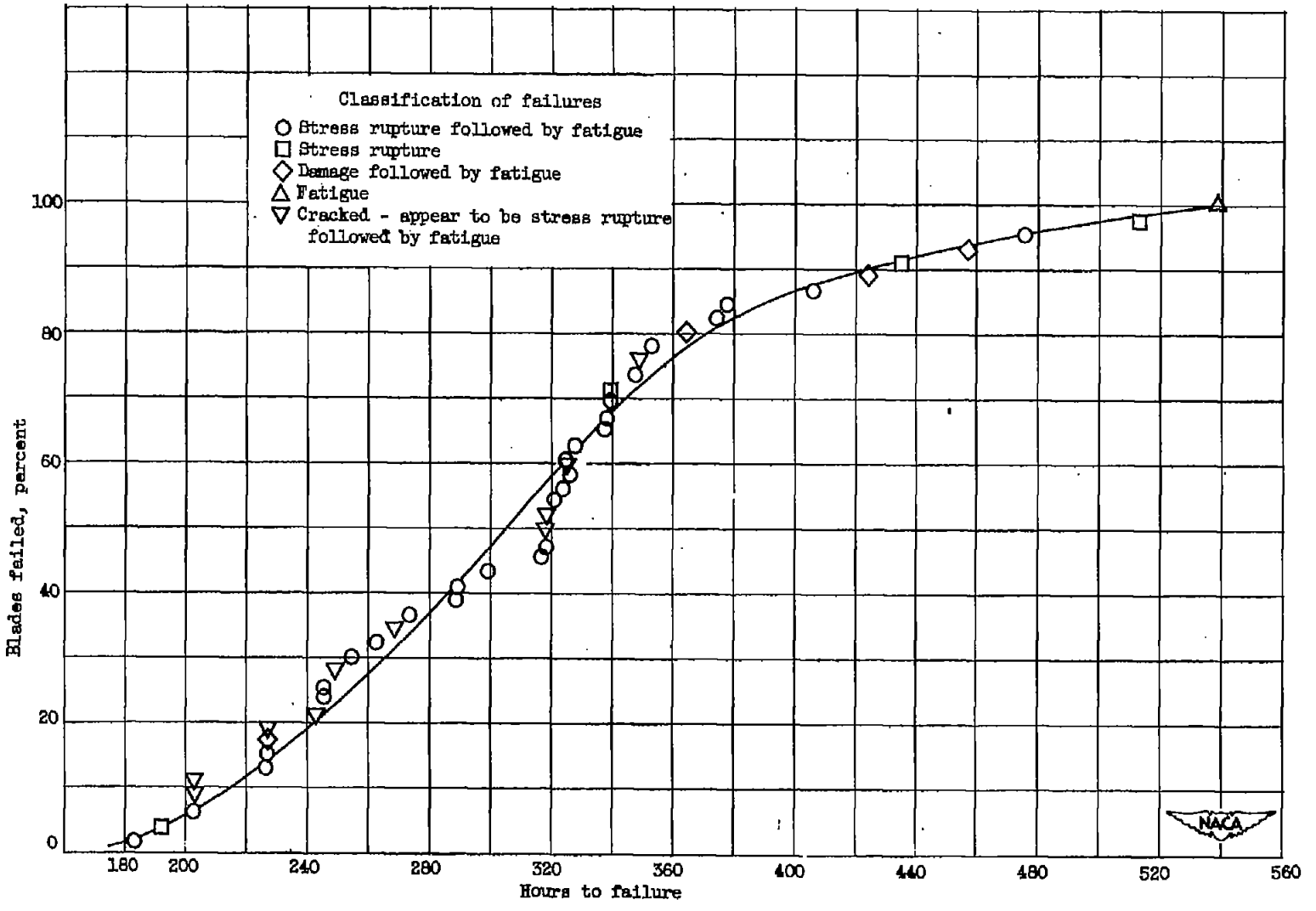
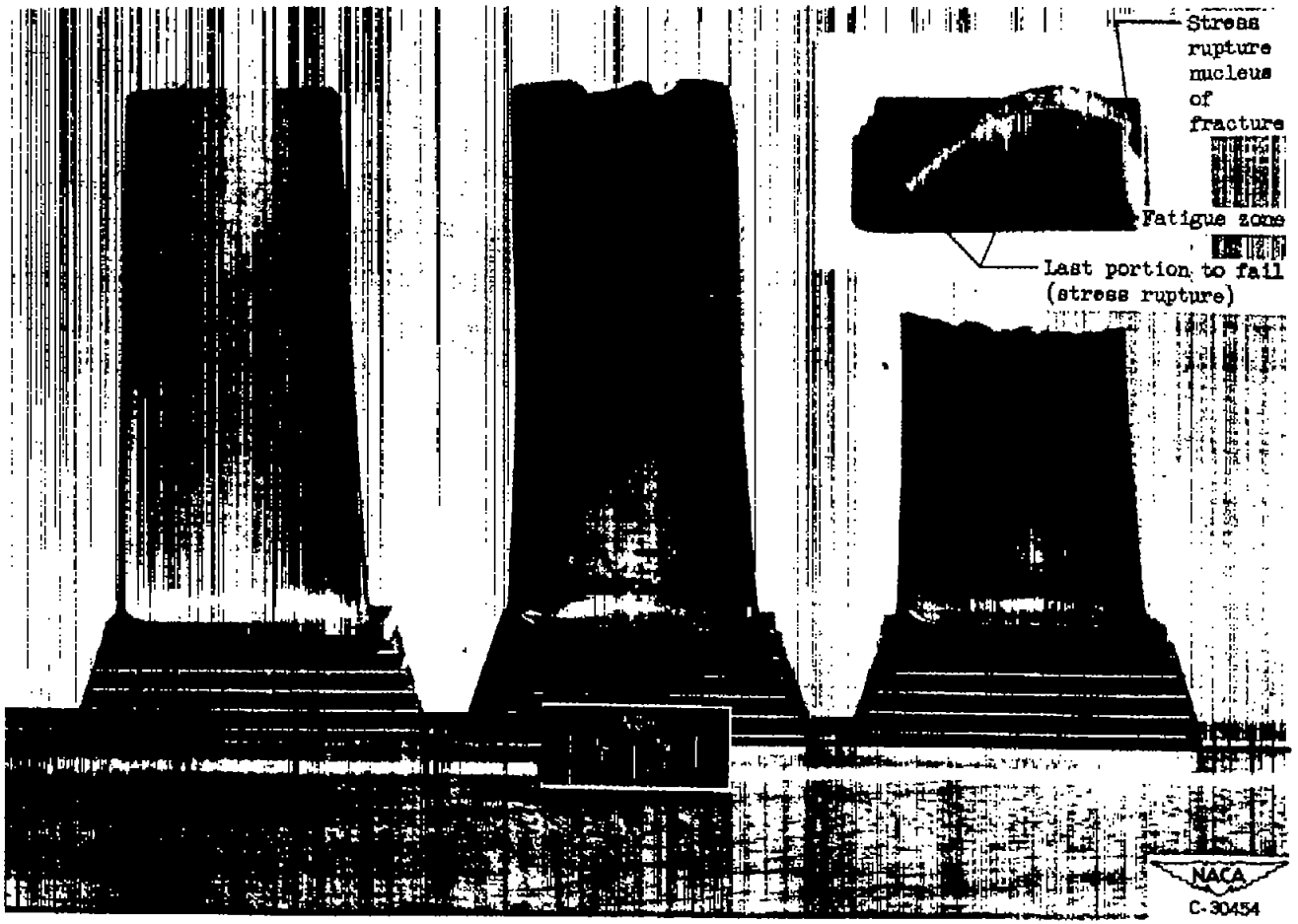


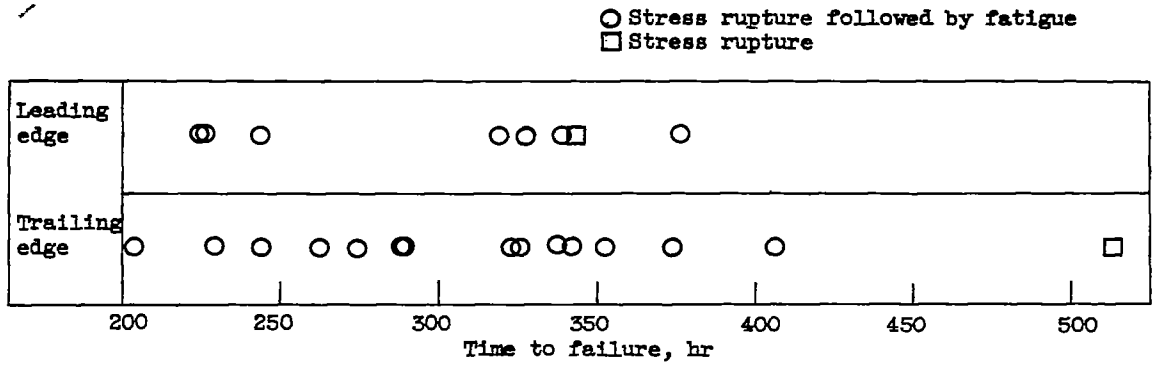
Figure 2. - Performance of forged S-816 blades in steady-state engine operation as shown by cumulative-frequency curve.



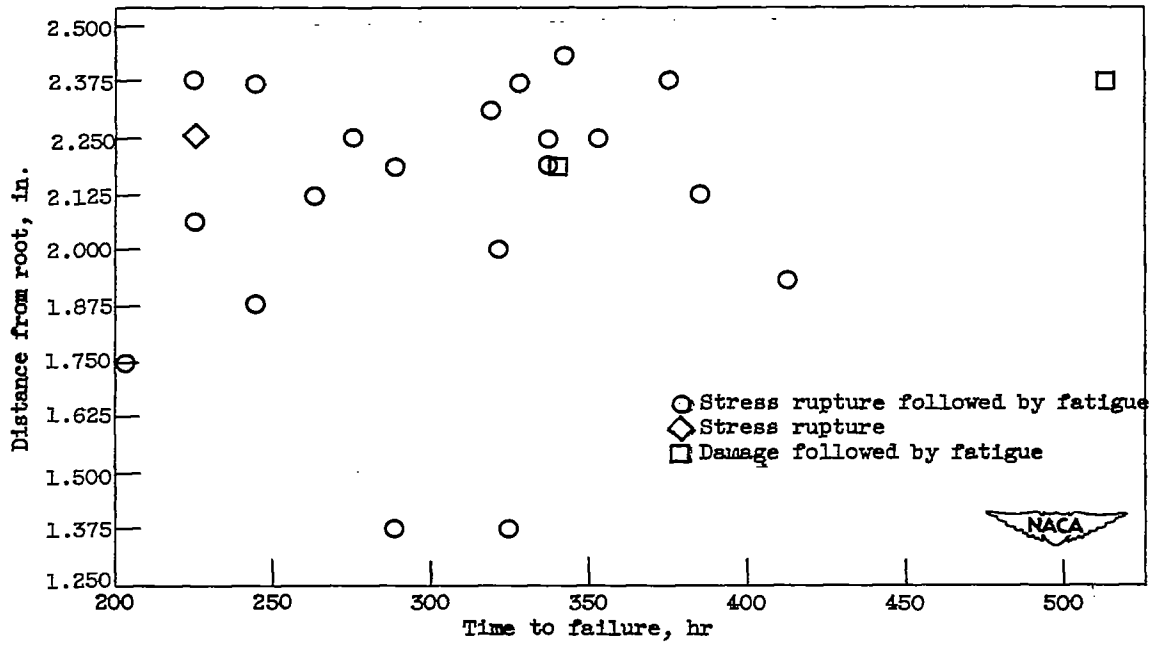
As-received blade Failure by cracking Fractured blade

Figure 3. - As-received blade and typical failed blades.

2745



(a) Origin of blade failures



(b) Origin of fracture edges relative to blade roots.

Figure 4. - Failure origins in blades.

[redacted]

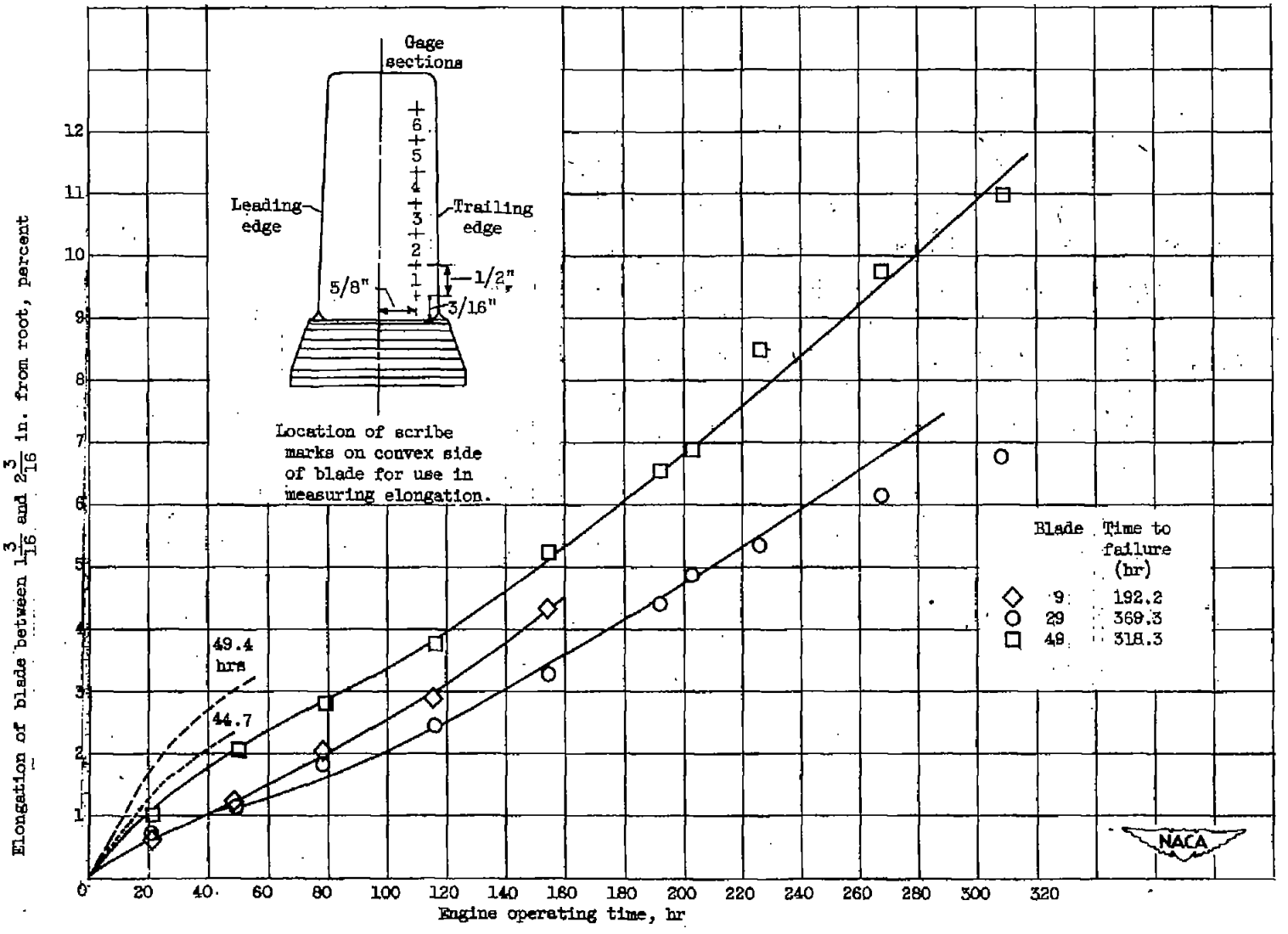
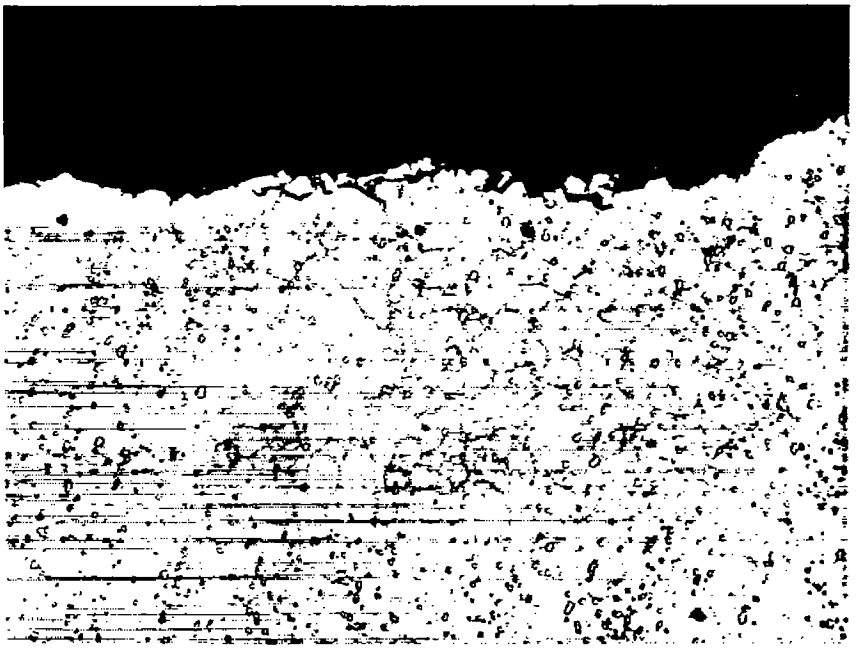
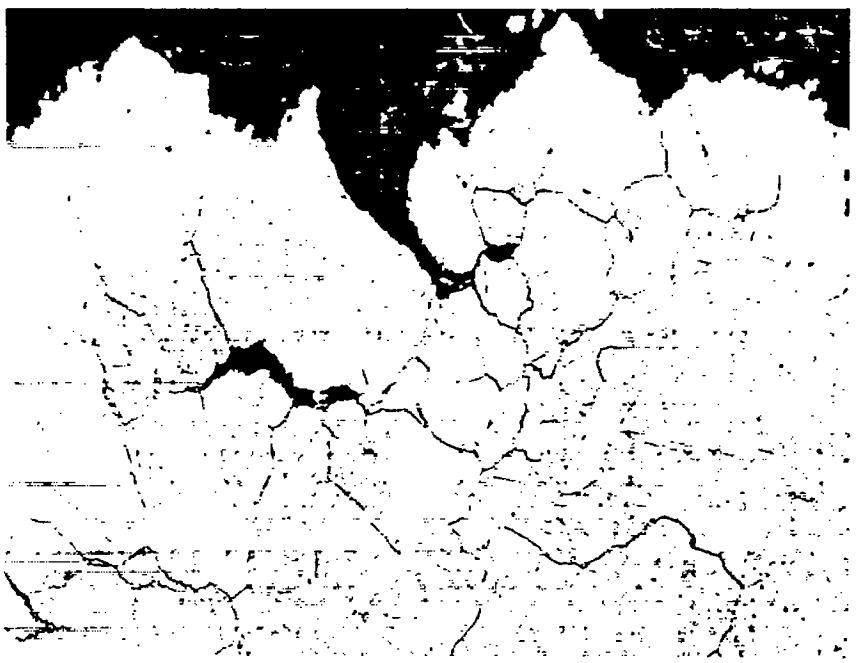


Figure 5. - Elongation of portions of blades during engine operation (gage sections 3 and 4). Dashed curves represent forged S-816 blades from reference 1. (Points shown on curves represent inspection periods.)

2745



(a) Microstructure near trailing edge of blade.

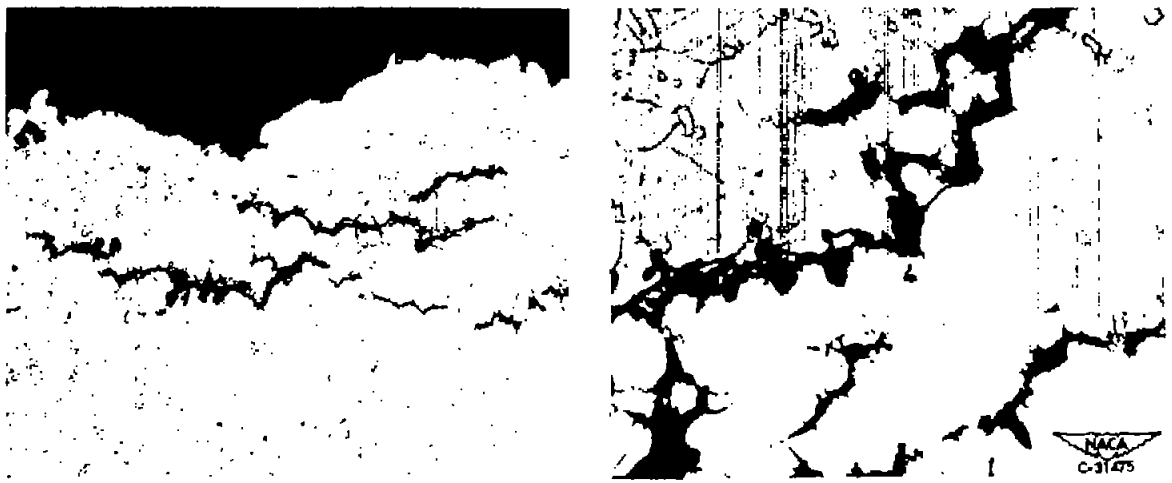


(b) Microstructure of area in center of blade.

Figure 6. - Microstructural differences in blades that failed in shortest time. (Failure time, 181.2 hr.) Etchant, 10 percent HCl; electrolytic; X250.

X250 X750

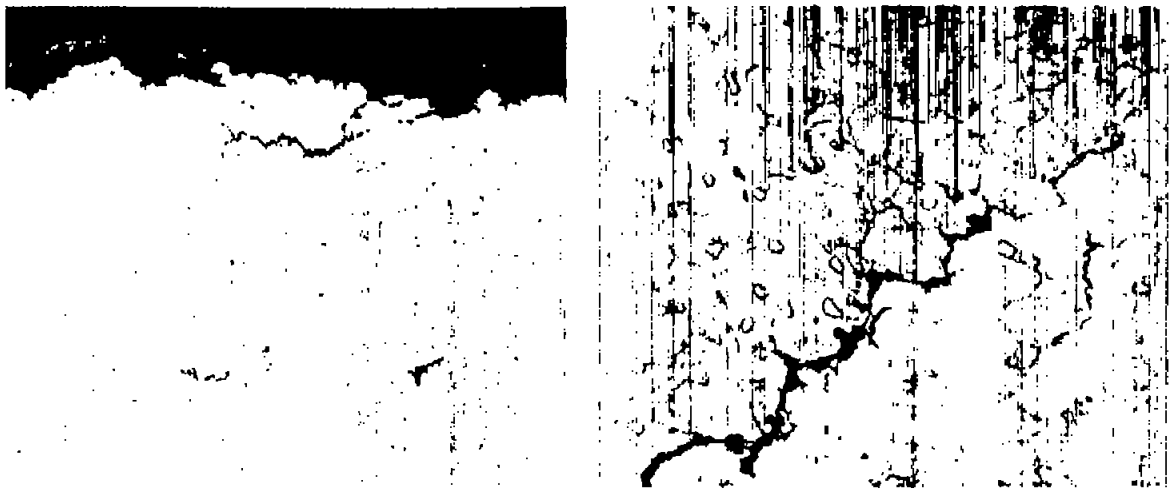
Fracture edge near center of blade 1. Failure time, 181.2 hours; grains coarsest here.



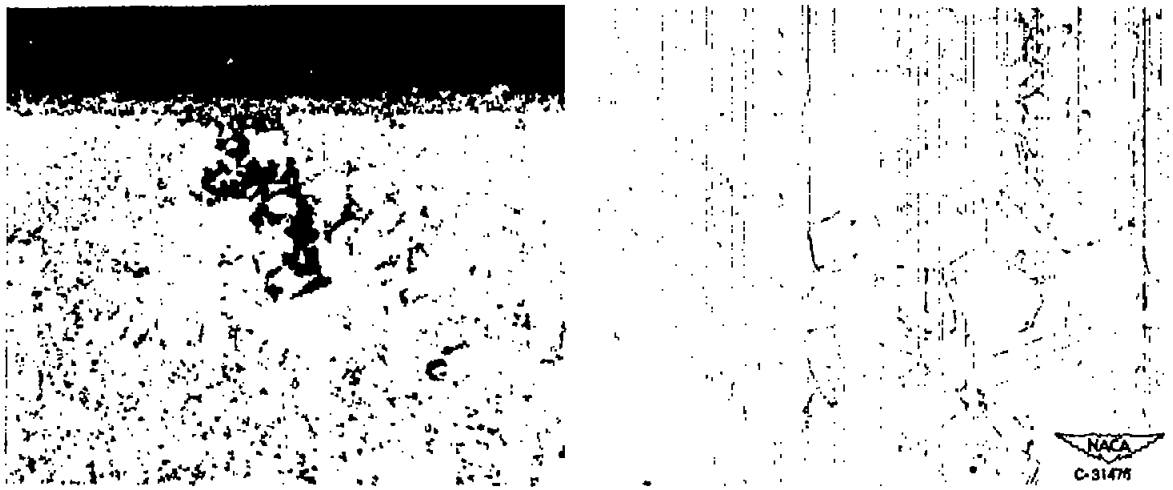
Fracture edge near leading edge of blade 3. Failure time, 254.6 hours.

Figure 7. - Microstructures of fractured or cracked blades at or near areas of failures. Etchant, 5 percent aqua regia in water followed by 10 percent hydrochloric acid in ethyl alcohol; both electrolytic.

X250 X750

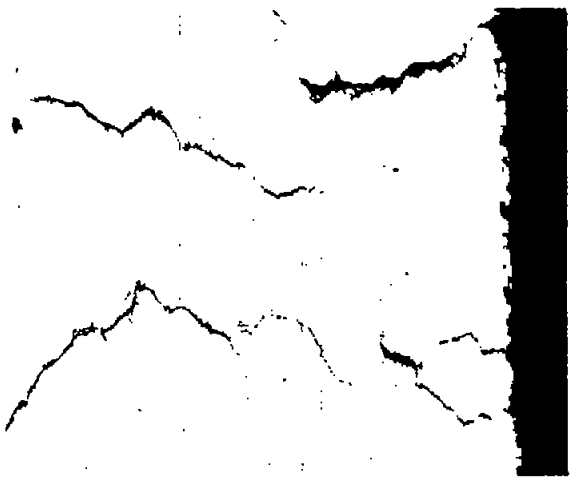


Fracture edge near trailing edge of blade 4. Failure time, 316.7 hours.

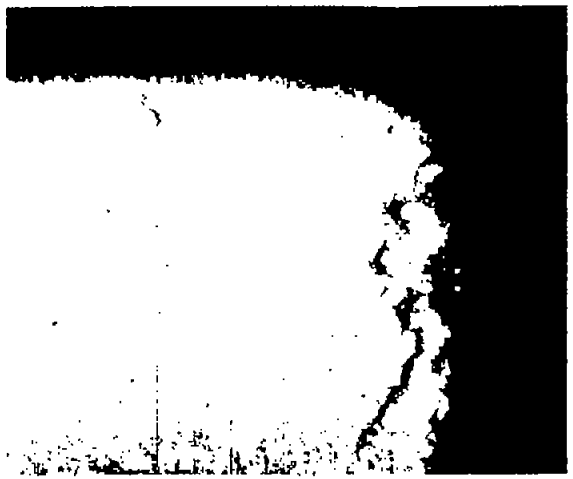


Crack near trailing edge of blade 5. Failure time, 434.2 hours.

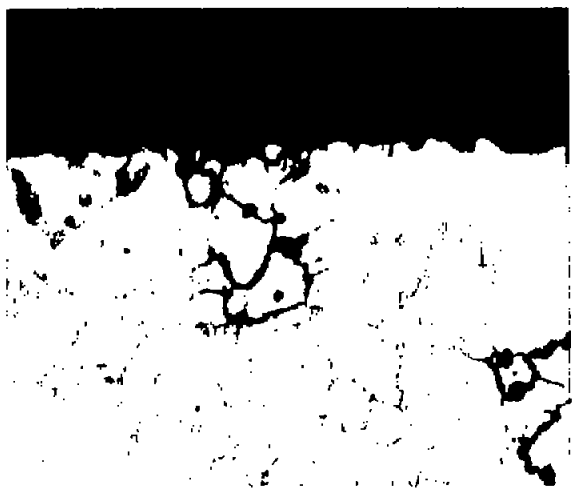
Figure 7. - Concluded. Microstructures of fractured or cracked blades at or near areas of failures. Etchant, 5 percent aqua regia in water followed by 10 percent hydrochloric acid in ethyl alcohol; both electrolytic.



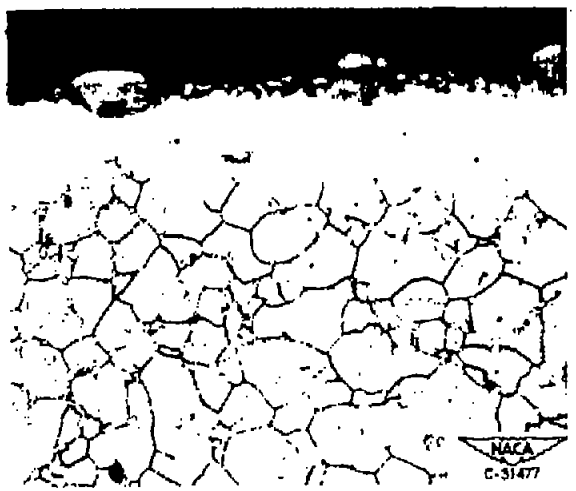
(a) Near fractured edge at center of air foil. X250.



(b) Fracture and trailing edges in fine grained, trailing edge portion of blade. X250.

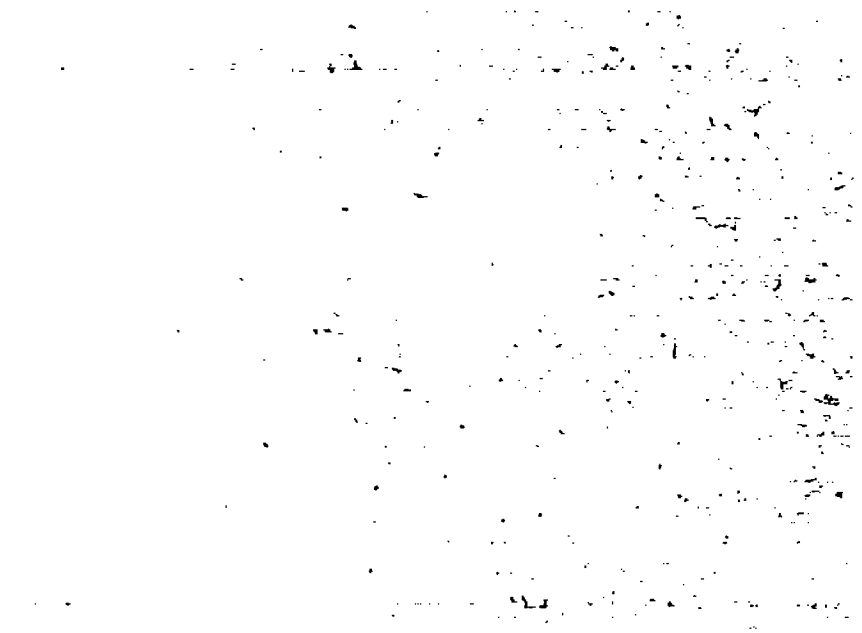


(c) Tip of trailing edge showing oxide penetration. X750.

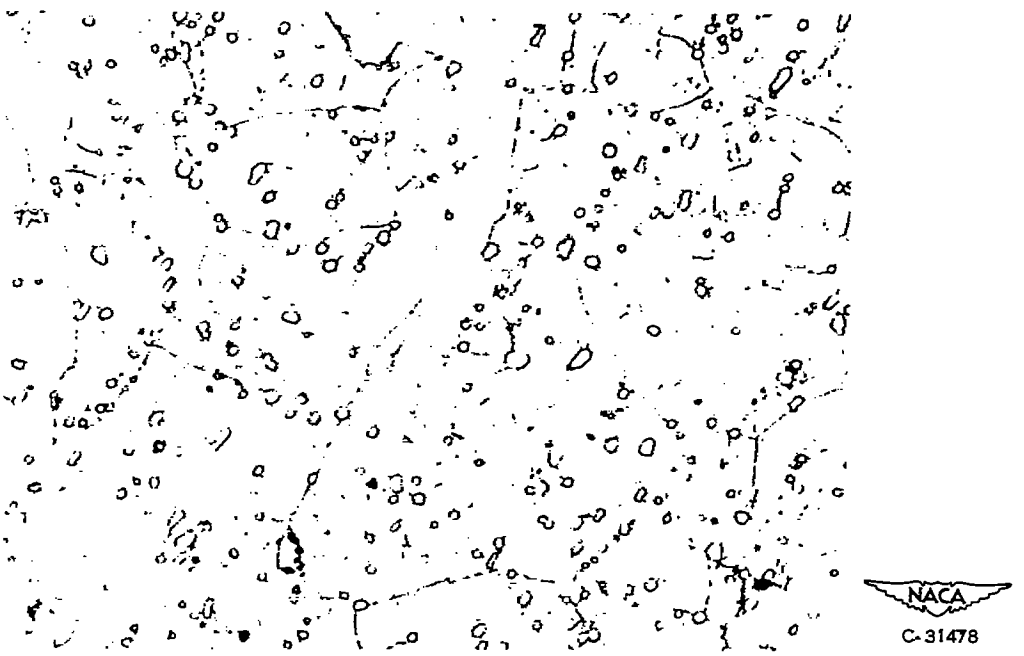


(d) Depletion zone on concave surface, (0.0012 in. deep). X750.

Figure 8. - Surface conditions present in blade that failed in 181.2 hours. Etchant, 10 percent HCl in ethyl alcohol; electrolytic.



(a) General structure near leading edge showing heterogeniety in grain size and large number of carbides per unit of area. X250.



(b) General structure near trailing edge showing salt and pepper precipitates, probably carbides. X750.

Figure 9. - Microstructure of older forged S-816 blades for comparison with structures of blades used in present investigation. Etchant, 5 percent aqua regia in water followed by 10 percent hydrochloric acid in ethyl alcohol; both electrolytic.

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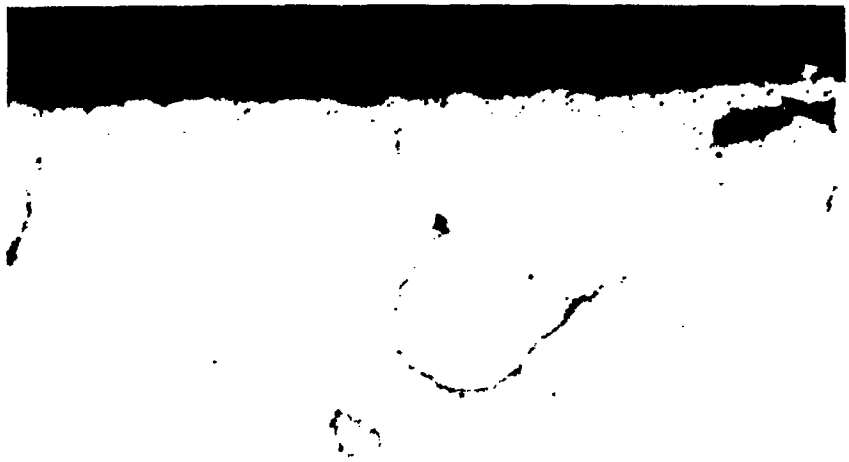
(a) Stress-rupture crack below actual fracture-edge; cross section of airfoil.



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(b) Probable oxide penetration near trailing edge on concave edge of airfoil.

Figure 10. - Edges of older blade run in steady-state operation for 78.9 hours before failure. Structures shown for comparative purposes.) Etchant, none. X250.

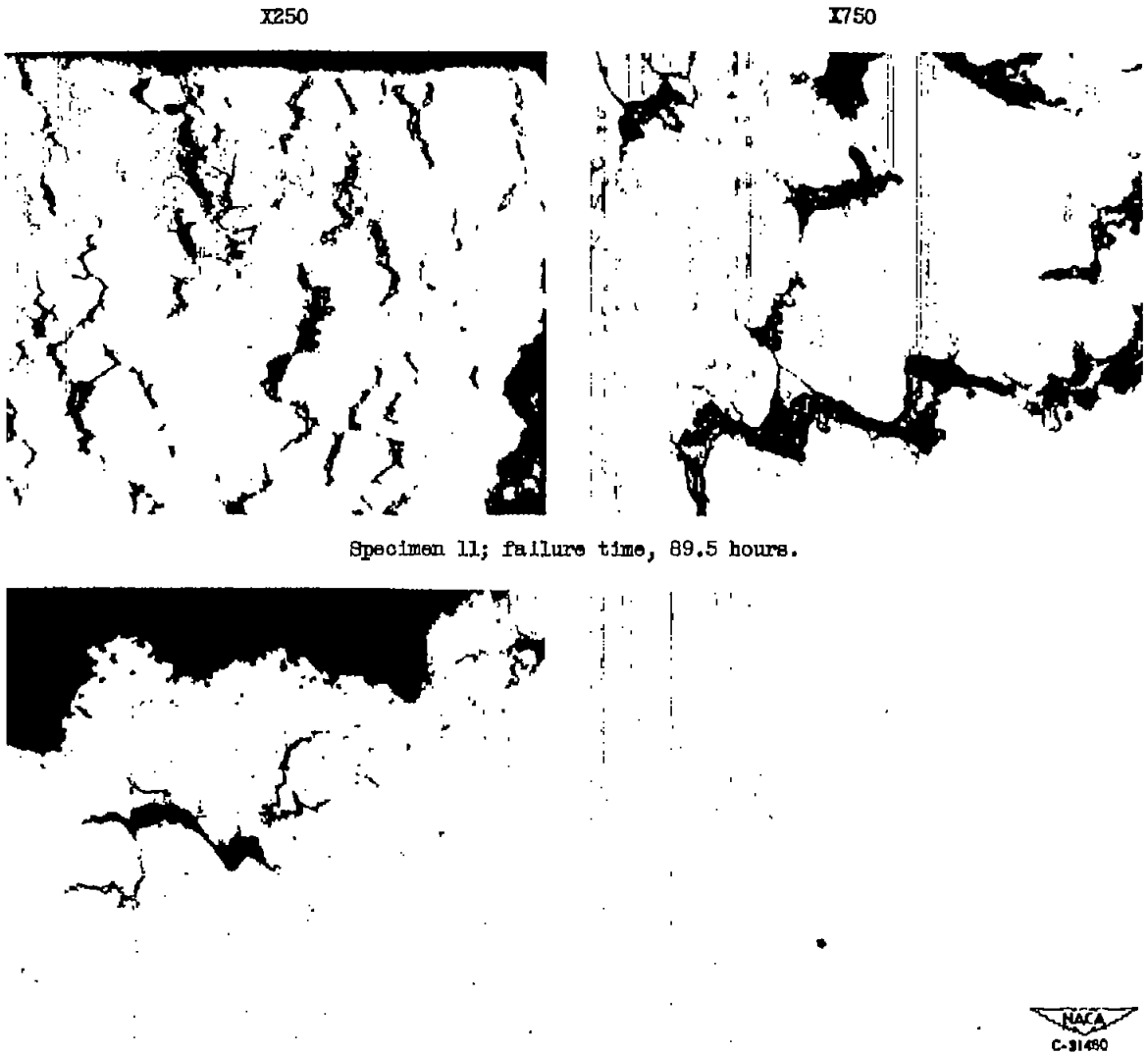
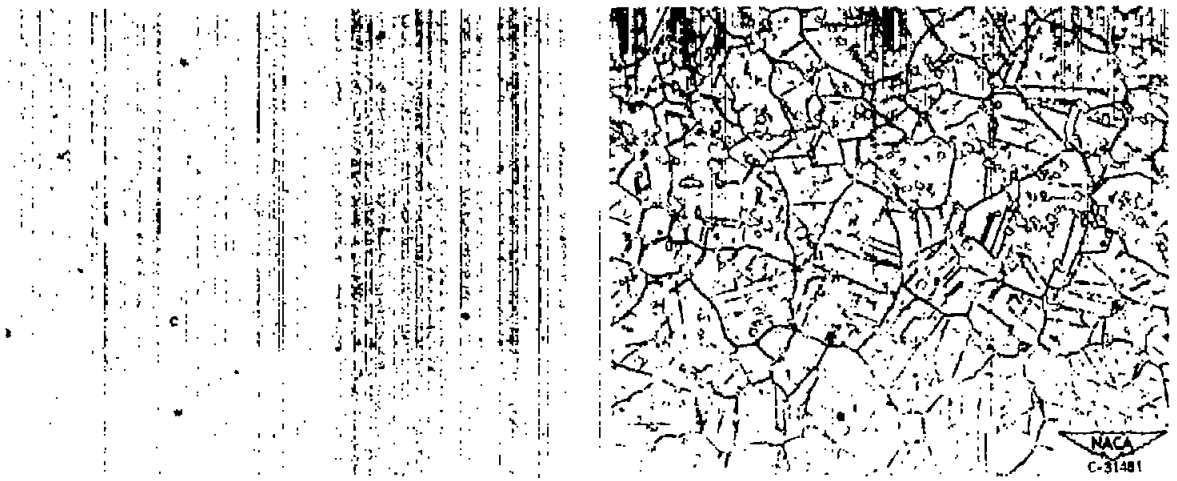


Figure 11. - Microstructures of fractured stress-rupture specimens showing intergranular tears and changes with time at temperature of 1500° F and stress of 20,000 pounds per square inch. Etchant, 5 percent aqua regia in water followed by 10 percent hydrochloric acid in ethyl alcohol; both electrolytic.

X250 X750

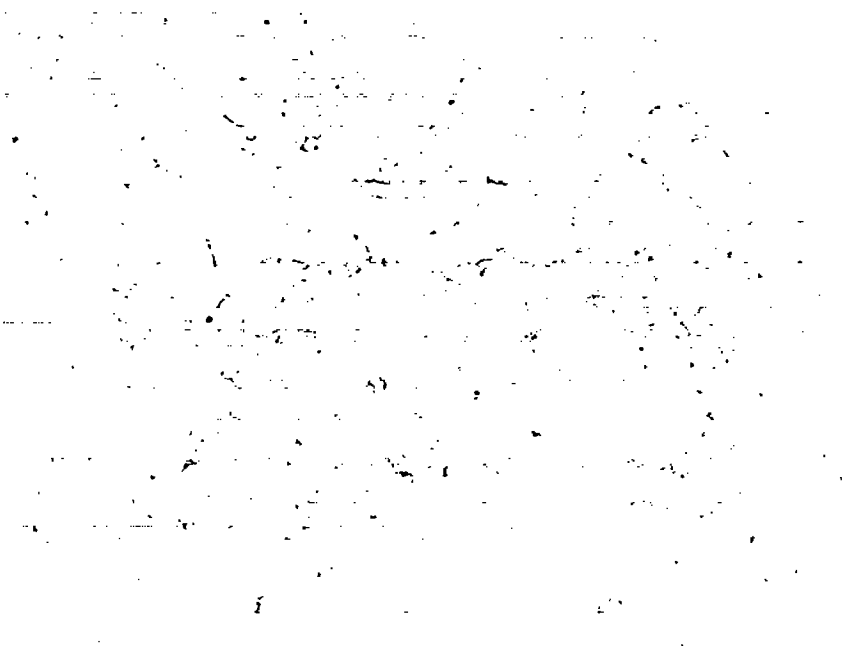
Specimen 6; failure time, 618.0



Specimen 1; failure time, 1346.3 hours.

Figure 11. - Concluded. Microstructures of fractured stress-rupture specimens showing intergranular tears and changes with time at temperature of 1500° F and stress of 20,000 pounds per square inch. Etchant, 5 percent aqua regia in water followed by 10 percent hydrochloric acid in ethyl alcohol; both electrolytic.

(a) Microstructure of untested segment of blade from which stress-rupture specimen that fractured in 89.5 hours was obtained.



(b) Microstructure of untested segment of blade from which stress-rupture specimen that did not fracture after 1346 hours was obtained.

Figure 12. - Differences in microstructures of untested segments of blades from which the poorest and best stress-rupture specimens were obtained. Etchant, 5 percent aqua regia in water; electrolytic. X250.

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